REVIEW PAPER

DISLOCATION INTERACTIONS WITH THERMAL PHONONS AND CONDUCTION ELECTRONS

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ABSTRACT

Experimental methods for the measurement of intrinsic interactions between moving dislocations and the crystal lattice are considered. It is emphasized that the stress pulse method is applicable at stress levels greater than about twice the static flow stress, while internal friction experiments may be used to explore the interaction at very low stress levels and small dislocation velocities. Recent results of low temperature stress pulse measurements in Cu are presented. The interactions deduced from measurements between 4.2^{o}K and 400^{o}K in some FCC metals are compared to theoretical predictions. Suggestions are made for future theoretical and experimental work on unresolved aspects of the intrinsic interactions.

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I. INTRODUCTION

The now classic work of Johnston and Gilman[1] in which dislocation velocities were measured as a function of stress in LiF lead the way to numerous direct measurements of the dynamics of dislocations in crystals. These measurements give information which is useful for predicting the deformation response of crystals, and for studying the origin and nature of the forces which impede dislocation motion. This paper examines the techniques used in experimental methods for the measurement of dynamic dislocation behavior. These methods fall into two categories: (i) indirect methods in which macroscopic parameters such as strain rate and stress or the energy loss in an ultrasonic stress wave are determined and (ii) direct methods in which dislocation velocities are deduced from dislocation displacement measurements in stress pulse tests or from observations of the time dependence of dislocation motion.

Experimental conditions for the determination of the strength of the interaction between a moving dislocation and an otherwise perfect lattice using the direct method are set forth. Theories of the temperature dependence of the interaction are discussed and compared to experimental measurements from $4.2^{\circ}K$ to $400^{\circ}K$. Unresolved aspects of the interaction between a moving dislocation and lattice phonons and conduction electrons are discussed.

II. INDIRECT AND DIRECT METHODS

Two indirect methods have been used for the measurement of the dynamic properties of dislocations: (i) measurements of internal friction and modulus decrement, and (ii) measurement of the strain rate vs. stress behavior at strain rates above about $10^2 \sec^{-1}$.

Dislocation displacements are of the order of atomic dimensions in the internal friction measurements, and dislocation velocities vary up to about $\pm$ 1cm/sec about a mean value of zero. The Granato-Lücke[2] string model is usually used in the interpretation of the data.

Considerably larger dislocation displacements and velocities are involved in the strain rate vs. stress experiments. In this type of test only the product of mobile dislocation density and average dislocation velocity may be determined. Hence a knowledge of the mobile dislocation density is needed to determine an average dislocation velocity uniquely. A reliable measurement of the mobile dislocation density cannot be made, so that reliable estimates of dislocation velocities are difficult to obtain from these experiments.

Since dislocation displacements are observed in the direct method, a theoretical model connecting dislocation dynamics with the measured parameters is not required. It must be recognized, however, that only an average velocity can be determined with this method. When the stress pulse technique is used and dislocation positions are determined before and after a rectangular stress pulse is applied, an average velocity given by the displacement divided by the pulse duration is found. When the dislocations are continuously observed during their motion, as in a transmission electron microscope, velocities are determined which are averages over the time period set by a video or motion picture framing rate or are averages over a displacement dictated by the resolution of the system used to observe the dislocation. Deviations from the average velocity which might take place as the displacement varies over atomic dimensions therefore cannot be detected. It is generally assumed that a terminal dislocation velocity is determined in the direct method and that the driving force supplied by the applied stress is equal to the retarding or drag force at that velocity.

We may divide the forces which impede dislocation motion into two categories, intrinsic and extrinsic. Intrinsic forces act on a dislocation in an otherwise perfect lattice while extrinsic forces are due to dislocation interactions with other lattice defects. Both intrinsic and extrinsic forces will act on dislocations under actual experimental conditions, since all crystals contain defects such as surfaces, vacancies, and impurities. Inertial forces are unimportant in most of the direct experiments.[3]

III. MEASUREMENT OF INTRINSIC INTERACTIONS

We consider here the problem of the measurement of intrinsic interactions in FCC crystals and on close-packed planes of HCP crystals. The motion of dislocations through an array of discrete obstacles has been treated theoretically by Kocks[4], Frost and Ashby[5] and others. It has been predicted[5] that local interactions with obstacles which obstruct dislocation motion become unimportant compared to velocity dependent intrinsic interactions at stress levels above about two times the critical stress required to drive the dislocation through the obstacles. At intermediate stress levels both intrinsic and extrinsic obstacles are effective; the stress-velocity relationship is non-linear and the total drag is greater than that due to intrinsic forces alone. These calculations were verified experimentally in zinc where the obstacle to basal dislocation motion was a forest of second order pyramidal dislocations.[6] We conclude from this that reliable direct measurements of the intrinsic interactions must be made at stress levels greater than two times the stress which causes dislocations to move macroscopic distances. In the internal friction experiments, the dislocations vibrate about their equilibrium position and are subjected to intrinsic forces. The stress levels are very low compared to the stresses required to overcome all of the extrinsic obstacles (the stronger extrinsic interactions fix the length of the vibrating segments and this effect is included in the theory). The intrinsic interactions may then be studied at either very low stress levels (internal friction measurements) or high stress levels (strain rate vs. stress and the direct method).

Extrinsic interactions are reduced by (i) minimization of impurities and the total dislocation density of the test crystals, (ii) using "fresh" dislocations which are free of segregated impurities and contain a minimum number of jogs, (iii) minimizing or avoiding interactions of dislocations on the active slip system or attraction of dislocations to free surfaces (these interactions may increase the driving force on the leading dislocations and cause the intrinsic drag forces to be underestimated). One method is described below which has been found effective in reducing extrinsic interactions so that the strength of intrinsic interactions could be determined. This method utilizes torsional stress waves to produce single, short duration stress pulses. Dislocation displacements are observed either on a cross section normal to the cylindrical specimen axis (the cross section is subjected to a stress distribution corresponding to that of static torsion) or on the cylindrical surfaces of the specimen (where the maximum torsional stress acts).

IV. TORSION TESTING

The method used to generate torsion waves with a $2\mu sec$ rise time has been described elsewhere.[7] A torsional stress state is advantageous in that microsecond duration elastic stress pulses are nondispersive in isotropic rods, in $\langle 100 \rangle$ axis rods of cubic crystals and in $[0001]$ axis rods of hexagonal crystals. A small amount of dispersion occurs in anisotropic cubic crystal rods with a $\langle 111 \rangle$ or $\langle 110 \rangle$ axis, but this dispersion is not a problem in the experiments.[8] Isolated dislocations are introduced for study by scratching[9] or by the use of line focused laser pulses.[10] The torsion waves are coupled to the test crystal by a bonding agent which has a liquid to glass transition temperature very near the test temperature. Suitable agents which do not cause dislocation displacements in the cooling and heating cycles have been found for test temperatures between $44^{o}K$ and $400^{o}K$. The bonding agent used at $44^{o}K$ is sufficiently strong at $4.2^{o}K$ to pass stress waves whose amplitude exceeds $10^{7} dyn/cm^{2}$. Below $44^{o}K$, differential thermal expansions cause sufficient stress to move dislocations near the bonded surface during the heating and cooling cycles. Therefore only the mobility of dislocations intersecting the cylindrical surfaces of the crystal is studied below 44^{o}K. Examples of isolated dislocations on the basal slip system of zinc are shown in Figs. 1 and 2.

The dislocations revealed in the Berg-Barrett topographs of Figs. 1 and 2 are between 1μ and 2μ below the surface. They were produced by scratching the (0001) surface with an alumina whisker (75mg load). The crystals were annealed prior to scratching so that all of the grown-in basal dislocations within about 20μ of the surface were removed by climb to the surface. After scratching, a set of dislocations is observed parallel to the scratch and extending about 70μ on each side of it. The dislocation spacing is about 10μ and their interaction at this spacing is very weak due to the presence of the nearby free surface. Figure 1 shows the dislocation configuration after application of a torsional stress pulse (the pulse was applied within 1 hour of scratching and its duration was about 20μsec). The continuous scratch to the right of the figure shows dislocation displacements which are directly proportional to the radial distance from the center of the cylindrical crystal. The torsional stresses vary linearly with radius, and the entire surface experienced the same duration of loading so the dislocation displacement vs. radius curve indicates directly the shape of the dislocation velocity vs. stress curve. Dislocation displacements from the discontinuous scratch to the left of Fig. 1 show the effect of pinning at the ends of each scratch segment. Residual curvature of the dislocations indicates a small friction stress. The effect of end pinning is strong in the low stress region near the crystal center and small in the high stress region where the displacements are essentially the same as those of the unpinned dislocations on the continuous scratch. Figure 2 shows that the pinning is strong enough to permit the dislocation to act as a Frank-Read source. Cross-slip of some screw oriented segments is indicated. (Observation of the critical stress for Frank-Read generation from pinned edge dislocation segments of known length has been used to determine the line energy of a screw dislocation.[11])

The theories discussed below indicate that the dominant intrinsic interactions take place very near to the core region of the dislocation. We conclude from this that a free surface parallel to the slip plane and about 1μ away from the dislocation core will have a small effect on the intrinsic drag since its effect on the dislocation strains and displacements near the core is small. Dislocation displacements near an end surface of a cylindrical crystal in a torsion pulse test should then be representative of the displacement of isolated dislocations in the bulk crystal.

Recently we have observed the mobility of edge dislocations in Cu at $4.2^{\circ}K$ using the crystal geometry shown in Fig. 3.* A torsion test at room temperature was made to compare the dislocation displacements with previous measurements of the displacement of isolated dislocations on a cross-section of a cylindrical $\langle 111 \rangle$ axis crystal. Good agreement was found, indicating that interactions between the leading dislocation in a slip band and those dislocations following it were unimportant. The results of torsion pulse tests on Cu at $4.2^{\circ}K$ shown in Figs. 4 and 5 indicate

* To be reported elsewhere.

velocities up to about 2×10^4 cm/sec. The displacement data shown in Fig. 5 is evidence that extrinsic interactions were important in two tests (at the lower stress levels) and unimportant in the test at the highest stress. We believe that interactions between scratch produced dislocations on different slip systems are the major extrinsic interactions in these tests. The viscous drag coefficient $B = \tau b/v$ where τ = resolved shear stress, b = Burgers vector and v = dislocation velocity was found to be 1.0×10^{-5} cgs. The edge dislocation drag coefficient in copper determined in torsion pulse tests is shown in Fig. 6 as a function of temperature.

V. THE MAGNITUDE OF DISLOCATION-PHONON AND DISLOCATION-ELECTRON INTERACTIONS IN METALS

In solid materials with a small Peierls' barrier in the dislocation slip planes, such as in the primary {111} planes of FCC metals and the basal plane of HCP zinc, lattice vibrations (termed thermal phonons) and conduction electrons are postulated to be responsible for the primary sources of damping of the motion of moving dislocations in an otherwise perfect lattice. Even in metals with large Peierls' barriers, emission and absorption of phonons is thought to dominate the energy-absorption process from moving dislocations in the moderate to high stress region.

These mechanisms have been described extensively in the literature, at least in the case of elastically isotropic crystals. Three fundamental processes which are applicable outside of the dislocation core have been postulated to determine the influence of phonons on the mobility of dislocations. They include (i) the interaction of the elastic strain field of a dislocation with thermal phonons, (ii) radiation of thermal energy from stress-induced oscillations of a dislocation, and (iii) influence of a discrete lattice in which the group velocity of thermal phonons varies as a decreasing function of their energy.

The first of the above three mechanisms has been treated by two approaches, strain field scattering[12] and phonon viscosity.[13,14] Each source of dislocation damping has a characteristic velocity and temperature dependence and is expected to dominate in a particular temperature, dislocation velocity region.[15] Both effects become vanishingly small as the temperature is decreased towards absolute zero. In the region of the Debye temperature, energy dissipation by phonon scattering exhibits a linear increase with temperature, while the phonon viscosity effect becomes temperature-independent. There is the additional small contribution from macroscopic thermoelastic damping.

The second of the above mechanisms is operative whenever a dislocation is accelerating. Three conditions under which this may occur are described below. The acceleration of a dislocation from some uniform velocity (including zero velocity) under the influence of an external stress field, either in the case of the free movement through the lattice or of its motion between obstacles in the crystal, is accompanied by the radiation of elastic energy.[17] A dislocation moving through a crystal lattice is constantly accelerated and decelerated by the oscillating stress field of the thermal phonons. This interaction, or 'flutter' mechanism[18-20] as it is commonly known, gives rise to the radiation of thermal energy. Arguments have been presented to prove that the flutter mechanism becomes negligible at very low temperatures.[19] As the Debye temperature is exceeded, the thermal phonon mean free path decreases to the order of atomic dimensions. In this limit a significant volume of the dislocation strain field cannot be coherently excited by a thermal phonon, and the flutter mechanism again becomes unimportant. In addition, for a moving dislocation, the Peierls' barrier presents an oscillating stress field, with attendant thermal energy radiation.[21] This barrier is especially important in BCC metals.

The third mechanism mentioned above has been more recently proposed[22, 23] where it has been shown that in a discrete lattice at 0^{o}K a moving dislocation radiates energy when its velocity exceeds the group velocity of thermal phonons moving in its direction of motion. The analysis has been extended to finite temperatures to include the effect of strain field scattering.[24]

Calculation of the magnitude of the interaction between conduction electrons and a moving dislocation in metals at low temperatures shown that a temperature independent dissipative force is exerted on the dislocation.[25] It is thought that the dissipative force has less relative significance at more elevated temperatures, where the phonon dissipative mechanisms increase in magnitude.

Considerable experimental evidence has been accumulated in the measurement of the magnitude of dislocation-phonon and dislocation-electron interactions. In the temperature range 4.2^{o}K to room temperature, dissipative forces, measured in direct mobility experiments where extrinsic effects were small, have been shown to increase linearly with dislocation velocity up to about ten percent of the shear wave velocity.[8, 10, 26-29] This is in accord with the linear velocity-stress relation predicted by the conduction electron and each thermal phonon damping mechanism.

In addition, results of both the direct experiments referenced above and indirect measurements (e.g. Ref. 30-32) from about $30^{\circ}K$ to room temperature, have exhibited an increasing dissipation with temperature, in qualitative agreement with the increasing influence of thermal phonons with higher temperature. (There appears to be reasonable correlation of results from direct measurements with those indirect measurements in which a relatively narrow peak in the attenuation versus frequency curve was recorded. It has been pointed out[33] that a wide peak may preclude treatment of the attenuation versus frequency data on the basis of the most simple Granato-Lücke model.) Much experimental evidence also indicates that as the Debye temperature (of the order of room temperature for many metals with close-packed structures) is approached, the rate of increase of the damping coefficient is lessened. In the only direct measurement made above room temperature,[8] the edge dislocation damping coefficient in copper was found to be independent of temperature between $296^{\circ}K$ and $373^{\circ}K$ (see Fig. 6). While no single mechanism has as yet been shown to be dominant over the extended range of temperature in which mobility measurements have been made, the trend in B at higher temperatures is qualitatively supportive of phonon viscosity in this region. (It is interesting to note that four independent estimates of B in copper from internal friction data at room temperature[30, 32-34] all fall within $1\text{-}2 \times 10^{-4}$ cgs.

In three of the four cases, additional data extended the possible limits of B to a somewhat wider range of values. This is in good agreement with $B \simeq 2 \times 10^{-4}$ cgs from direct measurements.)

Brailsford[15] has argued that the dominant Fourier components of a moving dislocation's displacement field dissipate energy through the process of strain field scattering, while the viscosity effect cannot be important. According to his calculation for strain field scattering the dislocation damping coefficient in copper at room temperature is given by $B(Cu) \simeq 3.9 \times 10^{-3}$ cgs, or more than one order of magnitude larger than that indicated by the accumulated data. Mason, who originally postulated that a moving dislocation dissipates energy through phonon viscosity, estimated the damping coefficient in copper to be $B(Cu) \simeq 2.5 \times 10^{-4}$ cgs at room temperature.[13] Mason's formula for B contained an inverse dependence on the square of an inner "cut-off" radius about the dislocation center, within which the strained material did not contribute to energy dissipation. This radius was only roughly estimated, and hence Mason's value of B must be considered to be only an order of magnitude calculation. The experimental observations are in reasonable agreement with the predictions of the phonon viscosity calculations above the Debye temperature.

It is generally agreed that in the liquid helium temperature range, conduction electrons provide the dominant means of energy absorption from a uniformly moving dislocation in a continuous lattice (discreteness of atomic positions is ignored). A refined calculation of the electronic damping coefficient for edge dislocations based on isotropic elasticity theory has been presented by Brailsford.[25]

Its magnitude is calculated to be 0.25×10^{-5} cgs for copper, 0.39×10^{-5} cgs for lead, and 0.76×10^{-5} cgs for aluminum. The theoretical values are smaller than values obtained experimentally by less than a factor of two for internal friction measurements in aluminum[32] and lead[31], and by a factor of four for direct mobility measurements in copper. It is as yet not clear whether closer agreement between experiment and theory can be obtained by further improvement of the experimental work or by extension of the theoretical analysis to include such effects as elastic anisotropy.

Results of direct mobility experiments by Weertman and co-workers[28, 29] indicate that below 30^{o}K, B in aluminum and in the normal state in lead rises dramatically, in stark contrast to the results discussed above. The discrepancy between the two sets of observations has not yet been explained although it is noted that in the most recently published work in aluminum,[29] B is estimated by averaging data points , rather than using the data of the maximum asymptote as is now commonly accepted.

The unresolved aspects of intrinsic dislocation interactions suggest work in the following areas:

(i) theoretical treatment of dislocation-phonon interactions which account for crystal anisotropy. This is especially important in the case of phonon viscosity where large errors may be introduced by the use of an averaged Gruneisen number rather than directionally dependent values.

(ii) numerical calculation of the damping coefficient derived from discrete lattice effects (phonon dispersion relation) at low temperatures.

(iii) experimental measurement of the damping coefficient of screw dislocations at $4.2^{\circ}K$. Conduction electrons do not interact significantly with screw dislocations because of the absence of a dilitational stress field (in the case of elastic isotropy).

(iv) further direct measurement of dislocation mobility in the normal and superconducting states of superconducting metals. The difference in energy dissipation between the two states provides a direct measure of conduction electron damping.

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FIGURE CAPTIONS

Fig. 1. Basal surface of a zinc test crystal after application of a torsional stress pulse showing dislocation displacements from both continuous (at right) and discontinuous scratches. The cylindrical axis of the crystal (and position of zero shear stress) is at the bottom center. The applied resolved shear stress did not exceed the breakaway stress for the discontinuous scratch segment closest to the cylindrical axis and its dislocation loops collapsed when the stress was removed.

Fig. 2. Basal cleavage surface of zinc with a 0.63mm scratch, after application of a 10μ sec stress pulse at 66°K.

Fig. 3. The test specimen orientation used in the copper mobility experiment at 4.2°K. The laterial $(\bar{1}1\bar{1})$ surface is shown with scratches parallel to a Burgers vector direction at intervals of about 4mm. A dislocation loop produced by the scratching process and lying in the $(\bar{1}11)$ slip plane with Burgers vector in the $[110]$ direction is shown schematically both before and following expansion by the applied torsion stress. The dislocation intersects the lateral $(\bar{1}1\bar{1})$ surface in an edge orientation.

Fig. 4. An etched area on a lateral $(\bar{1}1\bar{1})$ surface of a copper test crystal after testing in the torsion machine at $4.2^{o}K$. The few large etch pits indicate the low background dislocation density before the torsional stress was applied. The long vertical slip bands trace the movement of individual dislocations from the scratch at the bottom of the photograph. The scratch contained a high density of dislocation sources which induced the considerable amount of dislocation generation in the form of slip bands. From Test No. 4.

Fig. 5. Edge dislocation displacement in copper at $4.2^{o}K$ plotted as a function of the time integral of resolved shear stress.

Fig. 6. Dislocation damping coefficient plotted as a function of temperature for edge dislocations in copper.

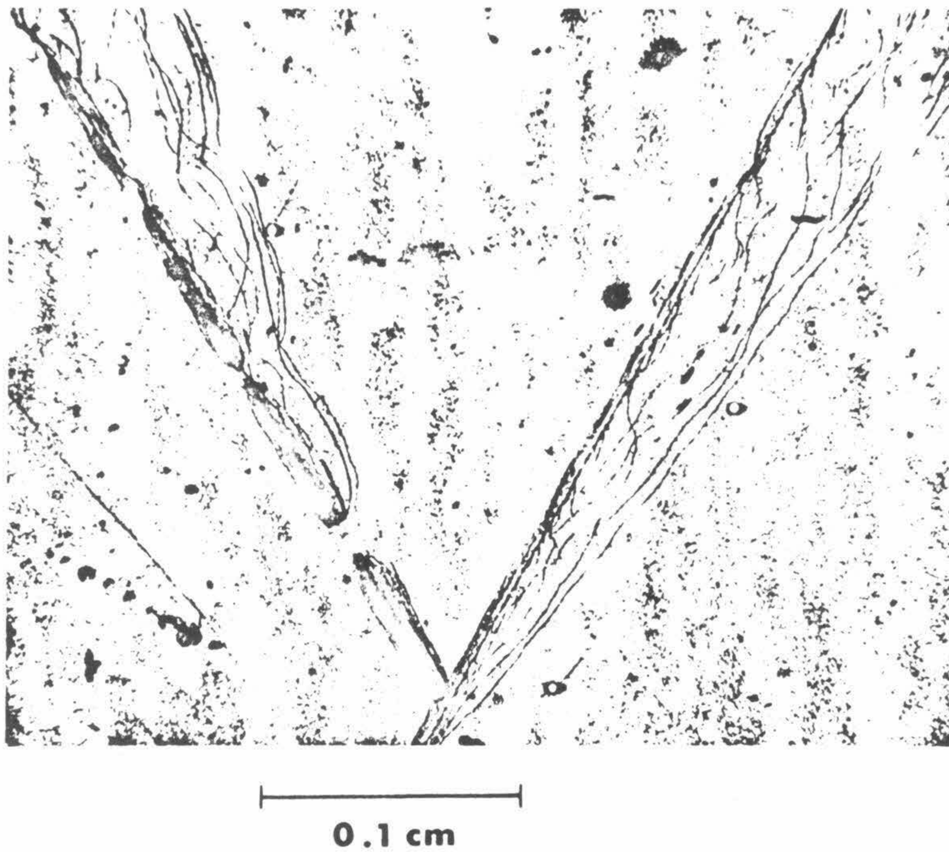
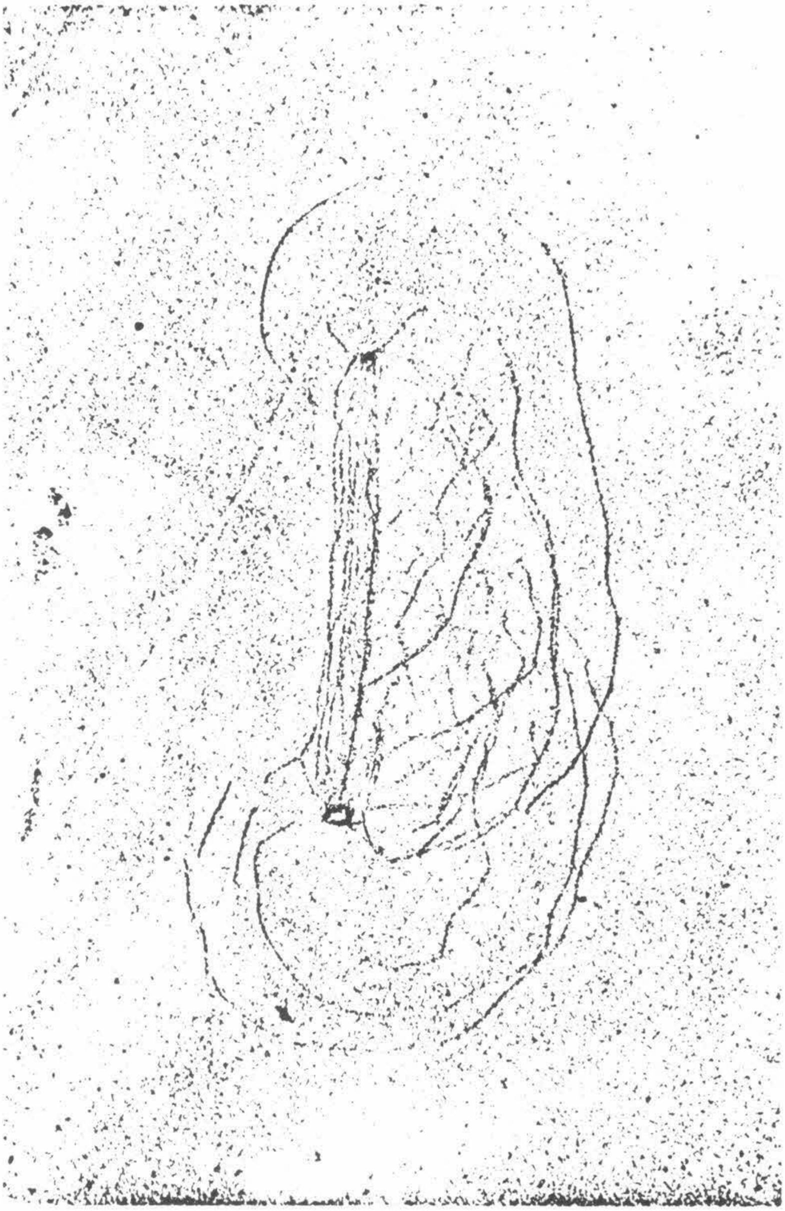


Figure 1



$\underline{b} = 1/3\ [1\bar{2}10]$

Figure 2

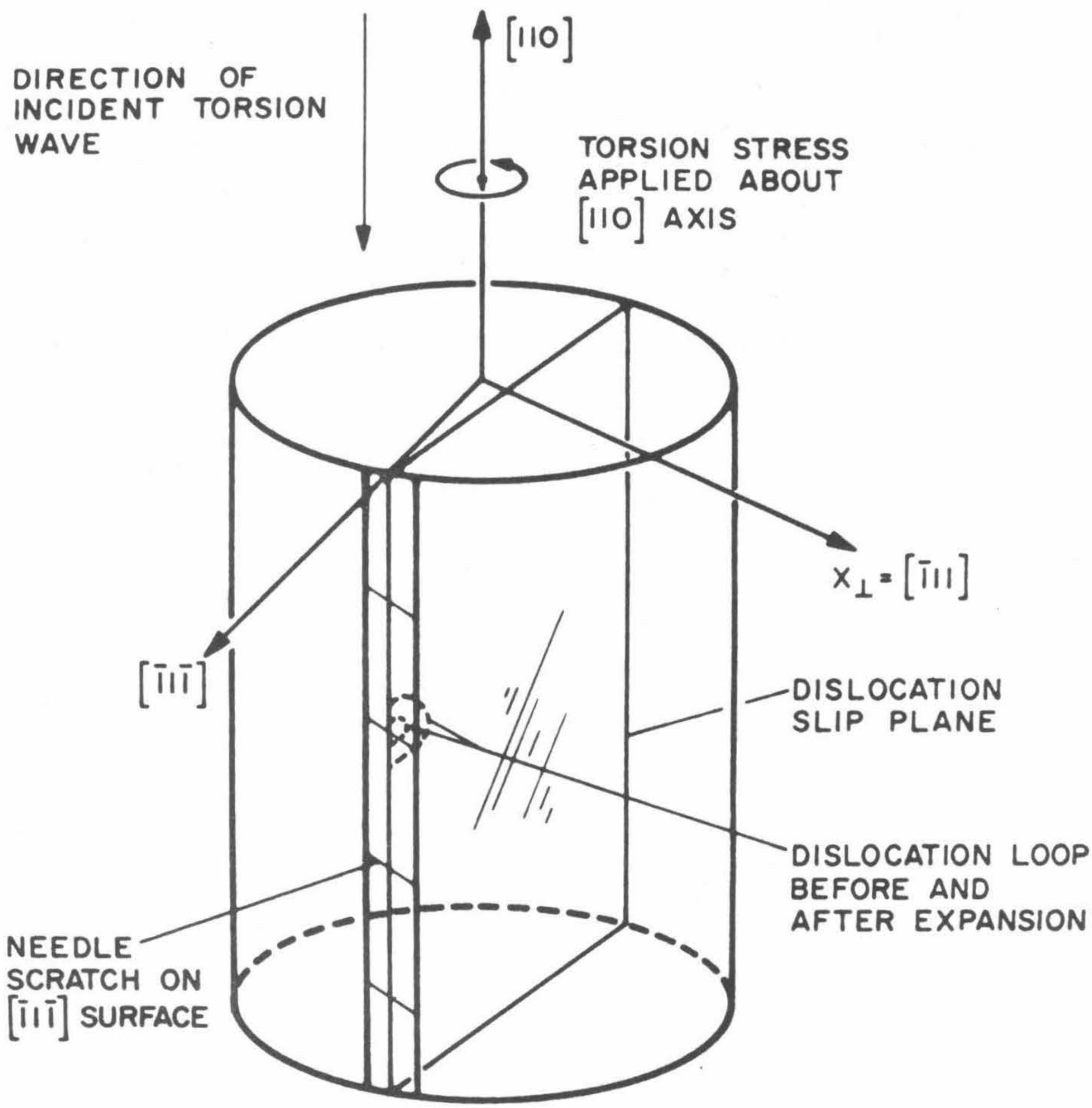

DIRECTION OF
INCIDENT TORSION
WAVE
[110]
TORSION STRESS
APPLIED ABOUT
[110] AXIS
$X_\perp = [\bar{1}11]$
$[\bar{1}1\bar{1}]$
DISLOCATION
SLIP PLANE
DISLOCATION LOOP
BEFORE AND
AFTER EXPANSION
NEEDLE
SCRATCH ON
$[\bar{1}1\bar{1}]$ SURFACE

Figure 3

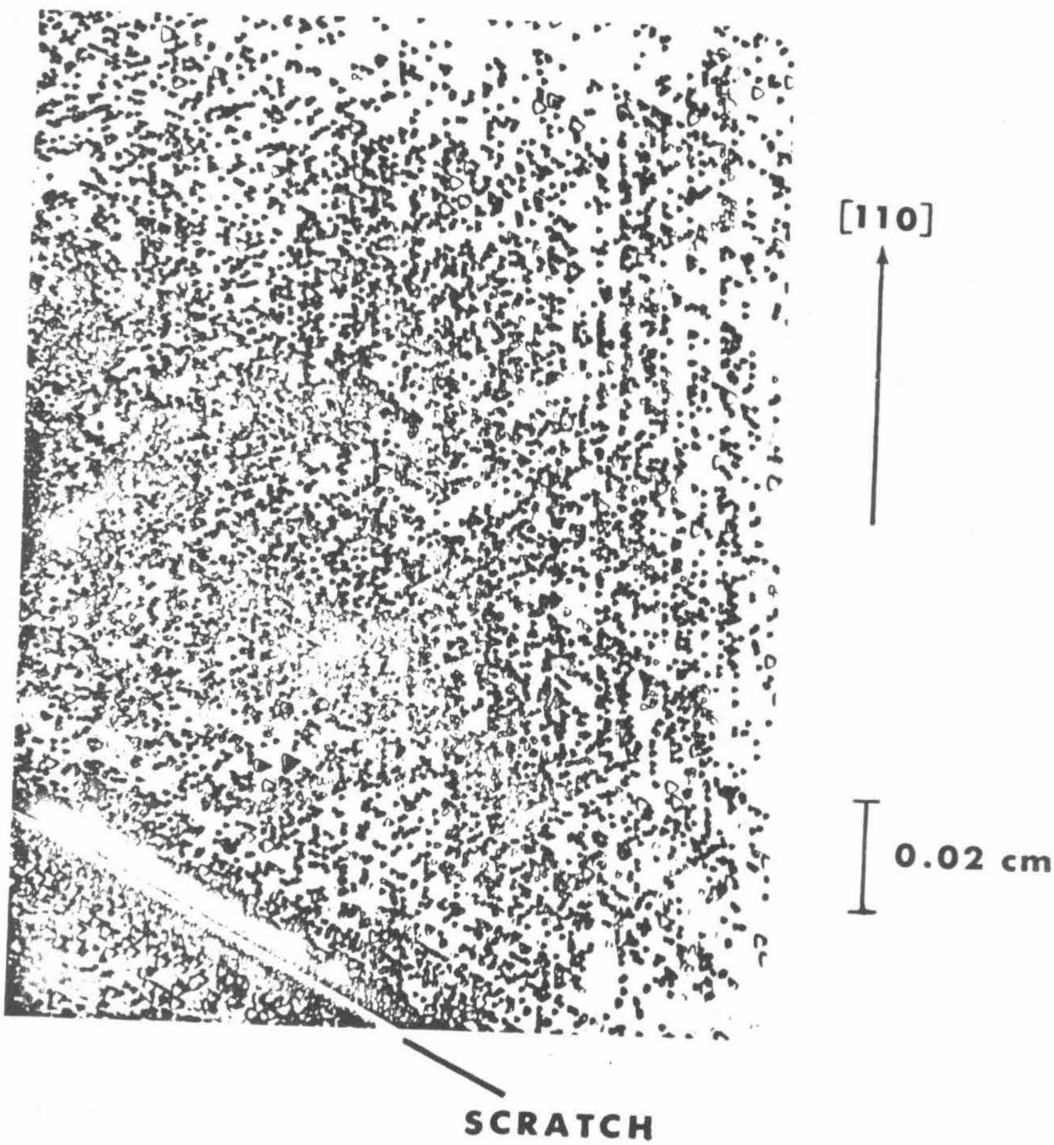
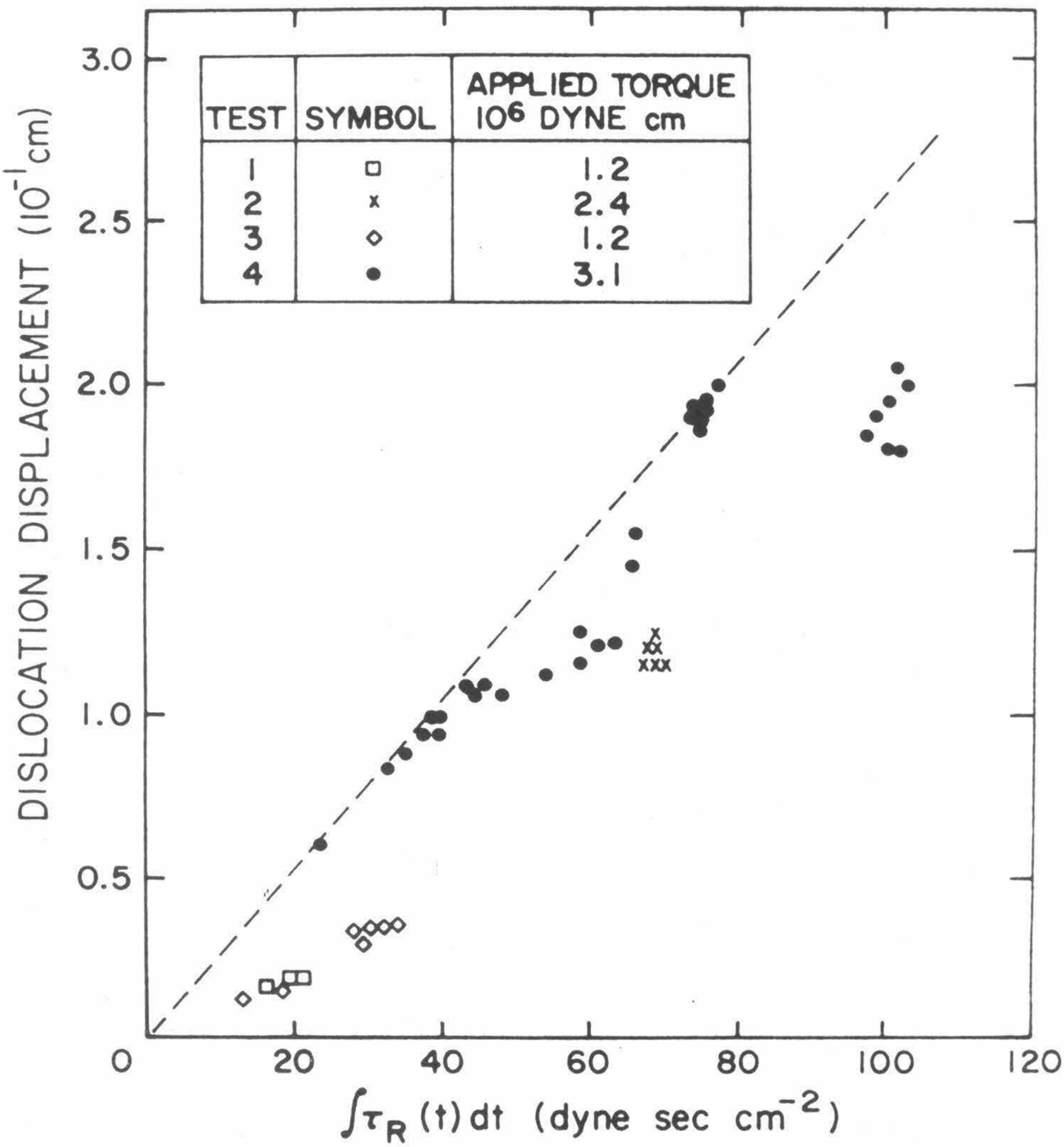


Figure 4



TEST
SYMBOL
APPLIED TORQUE
10⁶ DYNE cm
1 □ 1.2
2 x 2.4
3 ◇ 1.2
4 ● 3.1
DISLOCATION DISPLACEMENT (10⁻¹cm)
3.0
2.5
2.0
1.5
1.0
0.5
0
20
40
60
80
100
120
∫τR (t) dt (dyne sec cm⁻²)

Figure 5

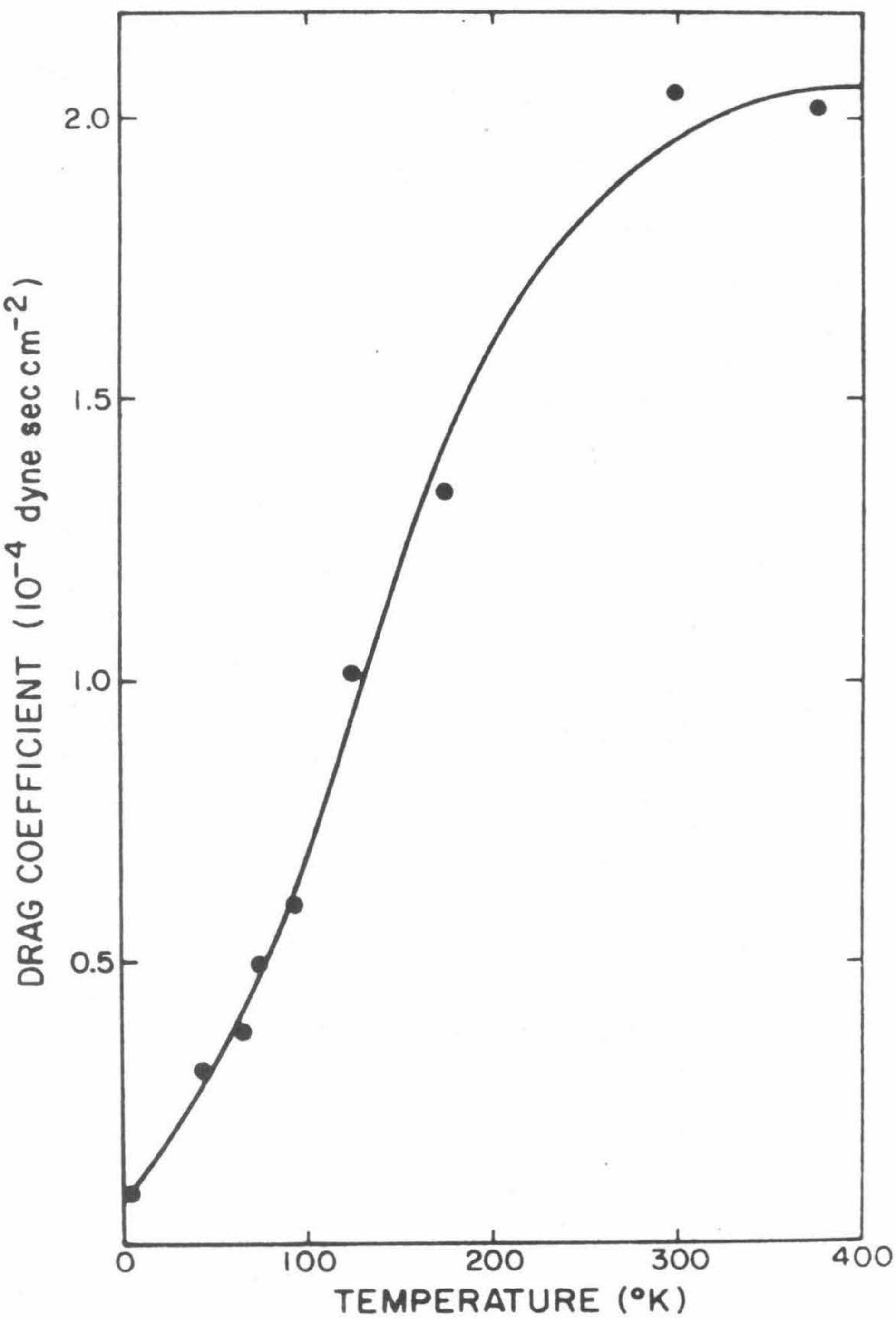

DRAG COEFFICIENT (10^{-4} dyne sec cm^{-2})
2.0
1.5
1.0
0.5
0
100
200
300
400
TEMPERATURE (°K)

Figure 6